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Initiation of stage I fatigue cracks – experiments and models

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Abstract

Early fatigue damage has been studied in a number of model and structural polycrystalline materials using modern experimental techniques. Early initiation of fatigue cracks was found to be related to the localization of the cyclic plastic strain into persistent slip bands during cyclic loading. Internal dislocation structure of the persistent slip bands and also the evolution of the surface relief as surface persistent slip markings are documented. The model of the surface relief formation describing the formation of extrusions and intrusions is briefly described. The intrusions representing sharp surface crack-like defects play the principal role in the initiation of fatigue cracks.

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1. Introduction

Appearance of fatigue fracture in crystalline material is the result of numerous processes taking place under the effect of external or internal variable forces. Variable forces induce variable stresses and variable strains arise. Provided the stresses are high enough the strains in the material are not only elastic but also plastic. Cyclic plastic straining results in irreversible changes in the material which are generally characterized as fatigue damage [1,2].

Plotting the stress vs. strain in a particular location of the structure during cyclic straining hysteresis loop is obtained. In sinusoidal or constant strain rate cyclic loading with constant stress or constant strain amplitude the

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shape and the magnitude of the hysteresis loop are changed only moderately during the fatigue life. In variable amplitude loading the stress-strain history is more complicated but we can identify the individual hysteresis loops using e.g. rain-flow procedure [3]. For most of the loading spectra the majority of the closed hysteresis loops will be elastic i.e. having zero width. Only if the width of the hysteresis loop is different from zero some damage is introduced in the material. Accumulation of the damage during the fatigue life leads to the initiation of a fatigue crack, or several cracks, their growth leading to final fracture.

It is the aim of this paper to clarify physical mechanisms leading to the initiation of fatigue cracks in constant amplitude loading and discuss the effect of variable amplitude loading on this process.

2. Cyclic strain localization in fatigue

Cyclic loading of material, either single crystals or polycrystals, results in appearance of cyclic strains in a particular location, usually in the area of stress concentration. Stress-strain relations in cyclic loading are characterized by the hysteresis loop. The width of the loop determines the applied plastic strain amplitude during one cycle. Plastic strain and also cyclic plastic strain in a material is facilitated by the presence of dislocations. Dislocation segments expand under external stress in a crystallographic slip plane and produce dislocation loops. Dislocation multiplication results in formation of slip lines on the specimen surface. Slip lines in uniaxial straining represent slip steps having small height. Therefore already in uniaxial straining plastic strain is not distributed homogeneously. Inhomogeneity of cyclic plastic strain is even more pronounced. In cyclic loading dislocations in active slip planes formed during initial tensile or compressive quarter cycle facilitate slip in opposite direction. Cyclic plastic strain is thus already early in fatigue life localized to very thin bands called persistent slip bands (PSBs). The volume of the PSBs represents a small fraction of the total volume of the material and therefore the local plastic strain amplitude is much higher than the average plastic strain amplitude applied to the material.

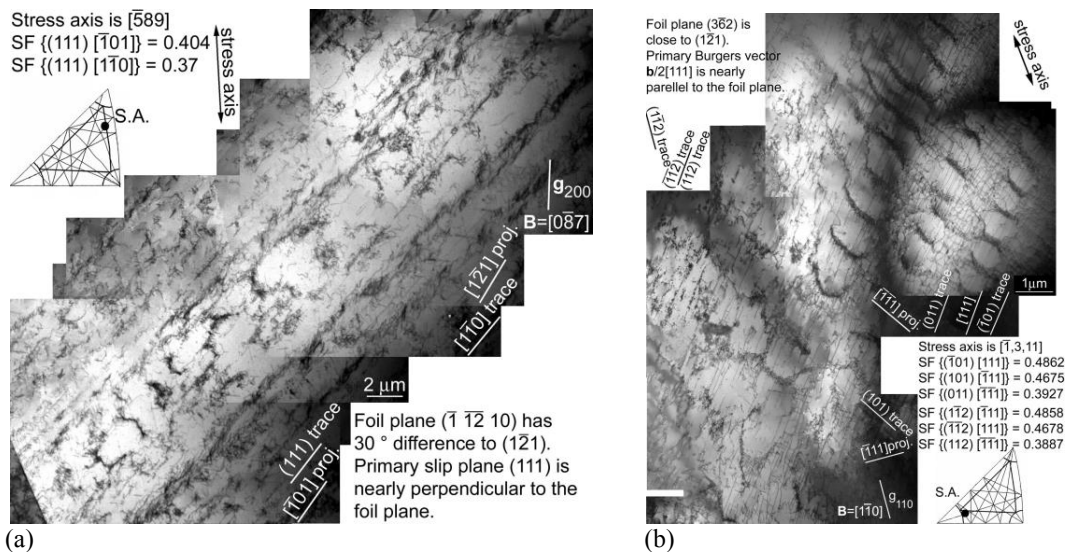


Fig. 1. Characteristic ladder-like dislocation structure of PSBs as revealed by TEM in (a) austenitic stainless 316L steel ($\varepsilon_{ap} = 1 \times 10^{-3}$, $N = 72000$ cycles) and (b) ferritic stainless steel ($\varepsilon_{ap} = 1 \times 10^{-3}$, $N = 9000$ cycles).

PSBs in the material represent thin lamellae whose thickness is around $1\ \mu\text{m}$ [4]. They run parallel to the dense crystallographic planes of a particular lattice. Fig. 1 shows the dislocation structure of two stainless steels fatigued with the constant plastic strain amplitude. Thin foil is oriented approximately perpendicular to the PSB lamella and parallel to the active Burgers vector of mobile dislocations.

Repeated cyclic plastic loading in the PSBs leads to the formation of pronounced surface relief in the form of specific slip markings in locations where PSBs egress on the surface of the material. Since these slip markings were formed during repeated cyclic straining they are called persistent slip markings (PSMs). Fig. 2 shows PSMs in a grain of austenitic stainless steel 316L cycled with constant strain amplitude to the early stage of fatigue life. Less developed PSMs consist of extrusions and central PSM reveals the presence of extrusion and parallel intrusion. Fig. 3 shows the AFM image which reveals the surface relief in three dimensions by direct observation of the surface (Fig. 3a) and using plastic replica (Fig. 3b). The intrusions are thin and often covered by an extrusion. Therefore only plastic replica (Fig. 3b) can reveal the presence of intrusions.

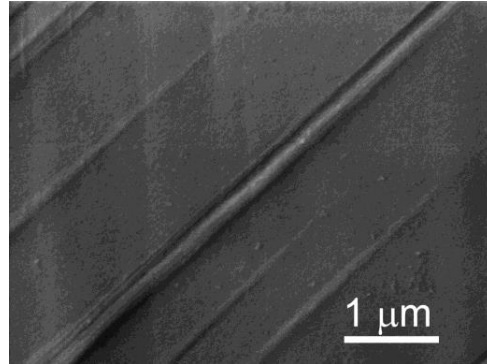


Fig. 2. SEM image of the grain of 316L steel cycled with $\varepsilon_a = 2.5 \times 10^{-3}$ for 500 cycles.

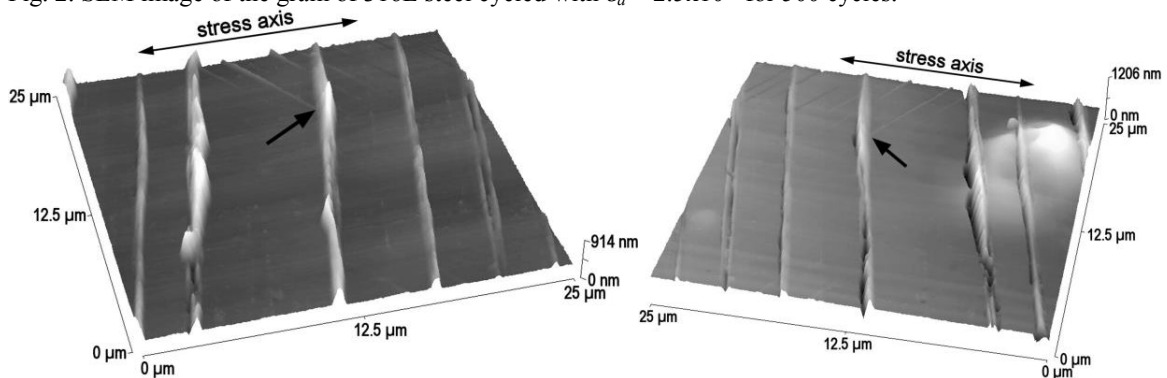


Fig. 3. Grain of fatigued austenitic steel cycled with $\varepsilon_{ap} = 2 \times 10^{-3}$ for 20 000 cycles, AFM; (a) metallic specimen, (b) plastic replica (inverted).

3. Locations of fatigue crack initiation

Since intrusions represent very sharp surface defects situated in the area of high local plastic strain amplitude fatigue cracks starts to initiate from the tip of the intrusion. Stage I crack is thus initiated [1] and starts to grow, initially along the PSB. The shape of the initiated crack under the specimen surface could be followed using focused ion (FIB) cuts [5]. Fig. 4a shows the surface of the grain of fatigued 316L steel with PSM whose surface length was around 24 μm . We have produced the FIB crater at the “sectioning start” (see Fig. 4a) and continued sectioning and photographing the perpendicular cuts. In Figs 4b - 4i several cuts corresponding to sections marked in Fig. 4a are shown. Fig. 4 shows that the crack started presumably in the middle of the grain where also the intrusion parallel to the extrusion was the deepest and from which fatigue crack developed.

The depth of the fatigue crack is largest in the middle of the grain (section 975, Fig 4i). The crack does not extend along the whole length of the PSM. Fig. 4b shows only the extrusion and a thin twin parallel to PSB. Fig. 4c reveals already a presence of a very short microcrack which further grows along the twin boundary. Detailed analysis of the cuts performed close to the centre of the PSM (sections 529 and 777) allows to identify the intrusion in the central part of the PSM (see Fig. 4f and Fig. 4g) from which fatigue crack presumably developed.

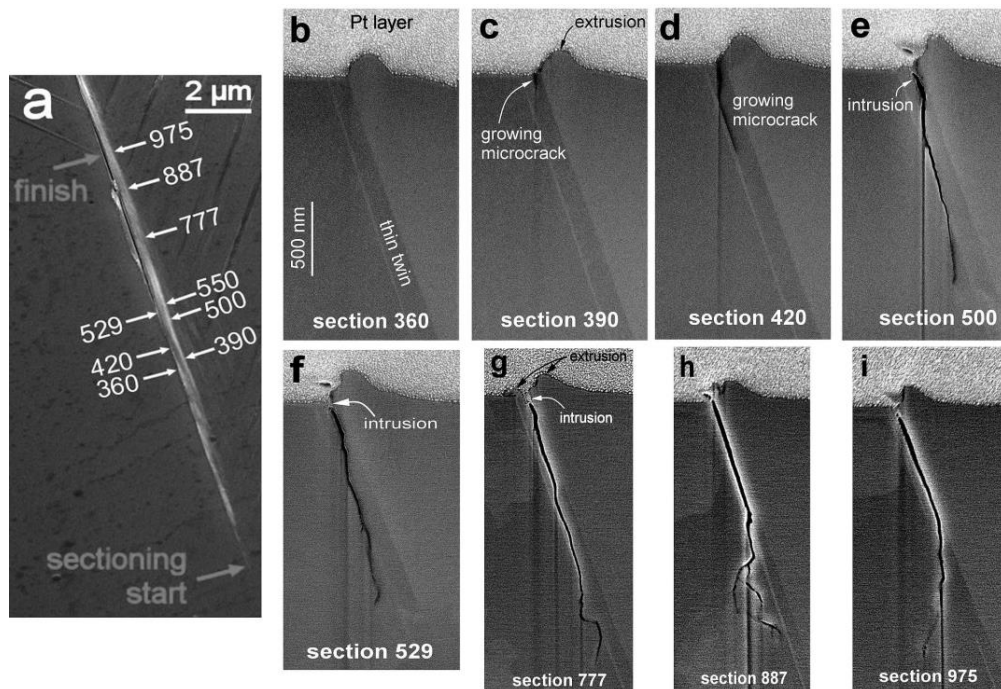


Fig. 4. PSM in fatigued 316L steel with primary crack (a) and FIB cuts (b) - (i).

4. Mechanisms and models of surface relief formation in localized cyclic straining

The mechanism of fatigue crack initiation has been an enigma since the early observations of the fatigue slip bands and cracks in Swedish iron by Ewing and Humphrey [6] at the beginning of the last century. Only in 70-ties TEM observations of the internal structure of fatigued metals led to the unveiling of the dislocation structure of PSBs and later to the demonstration of the cyclic strain localization. Numerous observations of PSMs in single and polycrystals revealed the preponderance of thicker extrusions but also the high incidence of thin intrusions (see [4,7]). It allowed formulating the physically based models of surface relief formation. Using the experimental data on the resistivity measurements indicating the production of not only dislocations but also point defects in cyclic straining Essmann et al. [8] proposed the model (EGM model) of extrusion formation due to point defect supersaturation in PSBs. EGM model, however, could not explain the steady growth of extrusions and the later the appearance of intrusions. Polák [9] extended the original EGM model considering not only the formation of point defects in PSB but also their continuous migration to the matrix. Quantitative description of the extrusion growth [10] based on Polák's model led to results in agreement with experiments. Only recently quantitative model of the formation of both extrusions and intrusions has been proposed by Polák and Man [11].

Quantitative model of surface relief formation in localized cyclic straining is based on the ladder-like arrangement of dislocations in PSB undergoing high plastic strain amplitude while neighboring matrix is cyclically

deformed only elastically. Mutual interaction of dislocations in PSB leads to the steady production of point defects with the rate p and their annihilation with the annihilation coefficient A . Fig. 5a shows schematically the section through PSB and the matrix and vacancies produced preferably in the channels.

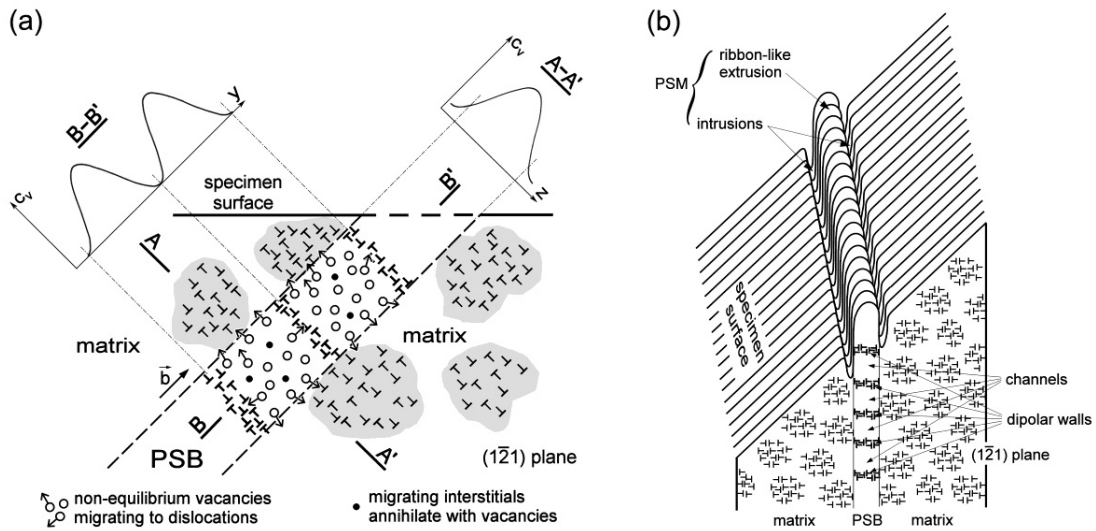


Fig. 5. Vacancy mechanisms of surface relief formation (a) schematics of ladder-like dislocation structure in a PSB and vacancy migration and annihilation at dislocations, (b) the shape of predicted surface relief at emerging PSB.

Systematic production of vacancies in the channels and their migration to the matrix represents the basic mechanism of transfer of matter between PSB and the matrix. Since vacancies migrate out of the PSB the atoms migrate from the matrix to PSB. The accumulation of matter in the PSB causes three-dimensional internal compression stress in the PSB. This stress can be plastically relaxed in the direction of the active Burgers vector and extrusion grows in the direction of the Burgers vector. Due to arrival of vacancies to the matrix the three-dimensional internal tensile stress arises in the matrix close to the PSB/matrix interface. The relaxation of this stress is more difficult than in PSB since matrix is harder than soft PSB. Therefore plastic relaxation starts later and intrusion on both sides of the PSB start growing with some delay relative to the growth of an extrusion. Fig 5b shows schematically the profile of the PSM consisting of a central extrusion and two parallel intrusions.

The shape of the PSM (extrusion and intrusions) can be obtained analytically [11] under some simplifying assumptions: (i) vacancies are produced in the whole PSB and all migrate to the neighbour matrix (ii) vacancies in the matrix are annihilated at homogeneously distributed edge dislocations having the density ρ_e . Provided w is the width of the PSB, l the depth of PSB under the surface, τ the cycle period and D_v the vacancy diffusion coefficient the vacancy concentration profile c_v vs. distance x from the centre of PSB can be obtained by solving diffusion equation under proper boundary conditions. In the PSB this equation is

$$\frac{d^2 c_v}{dx^2} - \frac{A}{\tau D_v} c_v + \frac{p}{\tau D_v} = 0 \quad \text{for } x \leq w/2 \quad (1)$$

In the matrix the equation has the form

$$\tau D_v \frac{\partial^2 c_v}{\partial x^2} - \tau D_v \rho_e c_v = 0 \quad \text{for } x \geq w/2 \quad (2)$$

Solving equations (1) and (2) under appropriate boundary conditions the vacancy profile c_v can be obtained. If the internal stresses in the PSB are fully relaxed due to cyclic creep processes and in the matrix only those above certain stress level, the growth rate of the extrusion height h_E (or intrusion depth d_I) can be found using relation

$$\frac{\Delta h}{\Delta N} = -\tau D_v \frac{d^2 c_v}{dx^2} \quad (3)$$

Performing the derivatives the height of the extrusion is proportional to the number of cycle N

$$h_E = \frac{plN \cosh(ax)}{\cosh(aw/2) + \frac{a}{\sqrt{\rho_e}} \sinh(aw/2)} \quad \text{for } x < w/2 \quad (4)$$

and the depth of the intrusion (d_I is positive)

$$d_I = r_{lm}N \exp\left(-\sqrt{\rho_e}\left(x - \frac{w}{2}\right)\right) - d_c \quad \text{if } r_{lm}N \exp\left(-\sqrt{\rho_e}\left(x - \frac{w}{2}\right)\right) > d_c \quad \text{for } x > w/2 \quad (5a)$$

$$= 0 \quad \text{if } r_{lm}N \exp\left(-\sqrt{\rho_e}\left(x - \frac{w}{2}\right)\right) \leq d_c \quad \text{for } x > w/2 \quad (5b)$$

where r_{lm} is the maximum intrusion growth rate at the PSB/matrix interface.

$$r_{lm} = \frac{pl \frac{\sqrt{\rho_e}}{a}}{\coth(aw/2) + \frac{a}{\sqrt{\rho_e}}} \quad (6)$$

and $a = \sqrt{A/\tau D_v}$ is the reciprocal characteristic diffusion distance which takes into account vacancy annihilation. d_c is parameter which characterizes the delay between the start of the extrusion and intrusion growth due to different yield stress of the PSB and the matrix. It is equal to the depth of a hypothetical intrusion provided yield stresses of the PSB and the matrix are the same.

From equation (5b) we can derive the critical number of cycles N_c below which no intrusion is produced and PSM consist only from an extrusion

$$N_c = \frac{d_c}{r_{lm}} \quad (7)$$

Eqs (5) and (6) were used to calculate the profile of the PSM consisting of a central extrusion and two parallel intrusions. Fig. 6 shows the profile of the PSM (only right part of a symmetrical profile) using typical set of parameters corresponding to cyclically deformed copper at room temperature.

Quantitative description of the shape of extrusions and parallel intrusions and their kinetics of growth was performed under several simplifying assumptions. The real dislocation arrangement in the PSB and in the neighbour matrix was idealized. The regular ladder-like dislocation arrangement of the PSB is not typical for all materials. However for steady production of point defects in the PSB and their migration to the matrix is enough if the dislocation free volumes alternate with dislocation rich volumes. It is typical for most materials subjected to localized cyclic straining. The production rate of vacancies p could be reduced and PSM builds up with smaller rate.

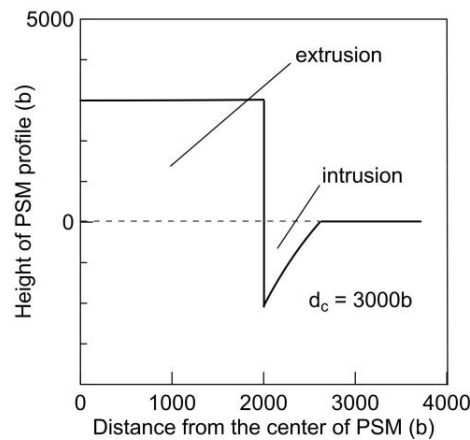


Fig. 6. PSM profile (right part of a symmetrical profile) consisting of extrusion and intrusion calculated using eq. (4) and (5). It corresponds to cyclic straining of copper at room temperature for 1000 cycles. The extrusion height in the centre is $3000b$, critical distance is $d_c = 3000b$. Edge dislocation density in the matrix $\rho_e = 1 \times 10^{-13} \text{ m}^{-2}$. All dimensions are in units of the Burgers vector modulus b .

More pronounced effect on the shape and the kinetics of the PSM has the assumption of the homogeneous distribution of edge dislocations in the matrix which serve as sinks for vacancies arriving from the PSB. Some typical distribution of dislocations in the matrix illustrates Fig. 1. In copper the matrix consists of patches and channels with approximately equal sizes. Contrary to the homogeneous dislocation structure the vacancies migrate larger distances from the PSB/matrix interface before they are annihilated at dislocations. It leads to slower intrusion growth. The local changes in the dislocation density of the matrix result in variable extrusion height and intrusion depth in agreement with experimental observations [12-14].

5. Discussion

Formation of the PSMs which contain intrusions provides the appearance of numerous sharp defects (crack-like defects) on the originally smooth surface of the material in areas with high local strain amplitude. The number of PSMs containing intrusions increases with the applied plastic strain amplitude. The depth of the intrusions increases linearly with the number of cycles. Simultaneously each of the growing crack-like defects concentrates increasing fraction of the local strain amplitude in the PSB into its tip. In the tip of the intrusion the large slip step representing a new surface is formed during tensile half-cycle. Due to the effect of environment the inverse slip step is not fully reversible due to incomplete rewelding of the new surfaces. From the tip of the growing intrusion in each cycle new surfaces are formed due to slip irreversibility. The whole defect could be considered as initiated fatigue crack when the increase of the depth of the defect due to local slip irreversibility is higher than due to the deepening of the intrusion.

The stage of fatigue crack initiation is important in the evaluation of the fatigue strength of materials and also for the prediction of the fatigue life though in high amplitude cyclic loading it represents a small fraction of the total

life. In variable amplitude loading the fatigue crack initiation stage can differ substantially for different loading spectra. Only several attempts were made to evaluate the effect of variable strain amplitudes on crack initiation. Lynn and DuQuesnay [15] used computer simulation to evaluate the effect of variable amplitudes on the crack initiation. Recently Li et al. [16] performed polycrystalline numerical simulation of variable amplitude loading effects on cyclic plasticity and microcrack initiation in austenitic steel. They have found the effect of overload and underload was different for two specific conditions.

The direct experimental investigations of the effect of variable loading conditions on the fatigue crack initiation are missing. The knowledge of the mechanisms of fatigue crack initiation and the kinetics of surface evolution (see sec. 3 to 5) indicate that the principal role in the formation of fatigue crack play cycles corresponding to closed hysteresis loops with the highest strain and stress amplitudes. Naturally, the effect of different strain amplitudes in the loading spectrum will also depend on the basic cyclic plastic behaviour of a material, namely whether material cyclically softens or hardens.

6. Conclusions

- (i) Experimental evidence reveals importance of the localized cyclic plastic straining in the formation of surface relief and initiation of fatigue cracks.
- (ii) Point defects play principal role in formation extrusions and intrusions on the surface of fatigued materials.
- (iii) Intrusions represent crack-like defects from which stage I cracks develop.
- (iv) Mechanism of fatigue crack initiation based on the physically founded model contributes to the understanding of the effect of variable amplitudes on crack initiation.

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